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(19)



Europäisches Patentamt  
European Patent Office  
Office européen des brevets



(11) Publication number:

**0 413 439 B1**

(12)

## EUROPEAN PATENT SPECIFICATION

- (43) Date of publication of patent specification: 30.11.94 (51) Int. Cl.<sup>5</sup>: **C22C 19/05**, C30B 11/00,  
C30B 29/52, C22F 1/10
- (21) Application number: 90307881.4
- (22) Date of filing: 19.07.90

(54) **Low carbon directional solidification alloy.**

(30) Priority: 14.08.89 US 393705

(43) Date of publication of application:  
20.02.91 Bulletin 91/08

(45) Publication of the grant of the patent:  
30.11.94 Bulletin 94/48

(84) Designated Contracting States:  
BE CH DE FR GB IT LI NL SE

(56) References cited:  
EP-A- 0 032 812      EP-A- 0 052 911  
EP-A- 0 150 917      EP-A- 0 225 837  
EP-A- 0 246 082      GB-A- 2 191 505

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**EP 0 413 439 B1**

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## Description

This invention relates to a directionally solidified (DS), columnar grain, low carbon, nickel based superalloy.

5 This alloy is designed to provide a high creep strength material capable of sustained operation at very high temperature, high stress conditions. It has been developed to provide a directionally solidified (DS), columnar grain alloy which imparts component creep performance and environmental properties (oxidation and hot corrosion resistance) similar to that of the first generation single crystal alloys. The alloy is designed for turbine engine airfoil components, the design of which makes them very difficult to be cast, or  
 10 at best, impractical to cast using single crystal technology. The size and shape of some components, particularly turbine vane segments and the long, low pressure (LP) turbine blades now being designed for advanced technology, high performance engines are not currently capable of being cast, using current single crystal technology. Therefore, it has become very important to develop a DS alloy having characteristics approaching those of the first generation single crystal alloys to allow for further turbine engine  
 15 performance improvements.

The new, more powerful, large, aircraft turbofan engines have increased diameter, and thus, increased length of the turbine blades. The LP turbine blade length  $\geq 200$ mm (8 inches) required of certain advanced design turbofan engines is such that they cannot readily be cast as a single crystal. Such blades encounter an unacceptable loss of thermal gradient during the single crystal casting operation and the resulting  
 20 castings begin to exhibit serious grain defects, such as moderate angle boundary defects which are referred to as "slivers", clusters of equiaxed grains known as "freckles", and high angle spurious grains. These grain defects particularly reduce the mechanical fatigue properties of the LP turbine blades.

A difficulty which this invention addresses is that the LP turbine blades for the new generation of very high performance aircraft turbofan engines require an alloy with creep strength in a temperature/stress  
 25 environment which heretofore has necessitated the use of a single crystal superalloy. However, at the same time, the performance requirements necessitate blade designs of a configuration which cannot be readily cast using the single crystal technique. When it became evident several years ago that advanced turbine engine designs appeared likely to encounter such a barrier for lack of a suitable alloy from which to cast the new LP turbine blades of long length  $\geq 200$ mm, the search for a new alloy was initiated. The starting point  
 30 was the high performance and proven, directional solidification, columnar grain nickel-base alloy CM 247 LC (US-A-4461659) because it is used to successfully cast the long LP blades and exhibits microstructural stability under high temperature/stress conditions. However, this alloy did not have the strength and creep resistance necessary for use under the temperature/stress conditions created by some of the newest, advanced turbofan engine designs. CM 247 LC requires a tightly controlled chemistry such that even small  
 35 variations from the narrow range of its acceptable composition can and do have marked and often unpredictable effects upon the alloy's performance characteristics. The greater the variation from the optimum chemistry the greater the adverse effect upon its performance characteristics. Therefore, the problem was how could its strength be increased to more clearly equal that of a single crystal alloy without having a serious deleterious effect upon its other desirable characteristics, such as reduction of its  
 40 resistance to grain boundary cracking during the DS casting operation, and increasing the alloy's tendency to form plate-like  $M_6C$  and TCP phase formation, among others. A further complication in the CM 247 LC chemistry as a base from which to work was the use of zirconium (Zr) as a creep strengthener. This presented a serious problem because whatever alloy was to be developed could not afford to lose the creep strength characteristics Zr provides in the CM 247 LC alloy, but at the same time, the tendency of Zr  
 45 to migrate to and concentrate at grain boundaries could lead to serious DS castability problems.

Another problem encountered in the attempts to successfully cast as a single crystal, parts such as turbine engine vane segments, is that of the residual stresses created by the difference in both the rate and the amount of thermal contraction between the metal of the casting and that of the ceramic mold into which the metal has been poured. These stresses occur as the casting solidifies and cools. This condition can  
 50 create excessive residual casting stresses. This residual stress can result in recrystallization during high temperature solution treatment of DS vane segments in alloys such as CM 247 LC. DS vane segments exhibiting recrystallization must be rejected due to the formation of undesirable transverse grain boundaries in the thin, upper or lower trailing edges of the vane airfoils. These transverse grain boundaries can nucleate thermal fatigue cracks during engine service. This situation is complicated and substantially  
 55 worsened when the design of the vane segment (see Fig. 9) has significant overhang at the top and bottom shroud, which tend to hang-up in the mold and increase the stresses within the airfoils resulting from solidification and cooling. At the present time, there is no known way of effectively relieving these stresses and achieving full gamma prime ( $\gamma'$ ) solutioning without occurrence of recrystallization.

These problems are addressed by the present invention according to which there is provided a low carbon nickel base alloy consisting of the elements set out in claim 1. The invention also includes a method of making parts of, components made from and use of the alloy in the way defined in claims 4-7.

Another important problem reduced by alloys of the invention is that of creep. The alloy of this invention is designed for use in very high performance engines. In application, the blades must be confined as closely as possible within their surrounding turbine casing to hold to a minimum gas leakage between the blade tips and the casing. Creep becomes a very important factor in limiting the degree to which the clearance between blade tip and the casing can be closed without causing excessive contact and the danger of turbine blade damage or failure. Additionally, modern turbine engine design can require the integrally shrouded vane segments support a bearing and hence the requirement for improved creep strength DS superalloys. When single crystal technology can be used, the problem of creep is materially reduced. One of the objectives of this invention is to approach the single crystal creep strength characteristic while using the columnar grain, directional solidification casting technique.

Because the application of the advanced vane segments and the low pressure blades require use of an alloy which could produce functional characteristics quite close to those of a single crystal alloy, it was necessary to add a strengthening element to partially compensate for the shift from single crystal to directional solidification casting techniques and to compensate for the necessity to eliminate high temperature solutioning processes to strengthen the alloy. For this purpose, rhenium (Re) was added. While Re increases strength, it also is prone to creating alloy phase instability. Re, under high temperature ( $>927^{\circ}\text{C}$  ( $1700^{\circ}\text{F}$ )) stress conditions can initiate the formation of plate-like, acicular appearing, topologically close packed (TCP) phases which are rhenium and tungsten rich and which if of sufficient size and frequency may nucleate premature mechanical fatigue cracks. The tests conducted on the alloy of this invention are evidence that the alloy composition herein disclosed has been able to take advantage of rhenium strengthening characteristics without the adverse effects resulting from creating excessive, large size and extensive rhenium, tungsten rich TCP phases. This has been accomplished by utilizing a very subtle balance of chemistry.

One of the advantages of single crystal superalloys is the ability to omit such elements as B, C, Zr and Hf because their basic purpose in high strength superalloys is that of strengthening grain boundaries. Omitting these elements increases the incipient melting point of the alloy. This has the advantage of permitting full solutioning of the '. This makes an important contribution to the creep strength of the alloy. Some advance design vane segment castings made from DS type alloys cannot be solution treated to this extent because there is the danger of recrystallization in the casting. Consequently, their thermal fatigue strength can be adversely affected. Thus, a different approach to obtaining strength improvement was necessary which is the purpose of adding Re.

The invention provides a directionally solidified, columnar grain, low carbon, nickel base alloy which derives its additional strength from the addition of rhenium and the substantial reduction of zirconium. Alloys according to the invention are essentially stable even after exposure at sustained temperatures above  $927^{\circ}\text{C}$  ( $1700^{\circ}\text{F}$ ).

Preferred features of the invention may be taken from the subsidiary claims which are also directed to components made from and uses of the alloys of the invention.

These and other advantages provided by this invention will be understood upon reading the following description and the accompanying drawings, in which:

Fig. 1 shows Larson-Miller curves of stress-rupture tests of DS longitudinal as-cast and double aged specimens comparing the alloy of this invention (CM 186 LC) with DS CM 247 LC "super solutioned" and double aged and CMSX-2/3 first generation single crystal nickel-base alloys;

Fig. 2 shows Larson-Miller curves of stress-rupture tests, DS transverse, comparing the alloy of this invention (CM 186 LC) as-cast and double aged, with DS CM 247 LC "super solutioned" and single aged;

Fig. 3 shows Larson-Miller curves of DS longitudinal, stress to one percent creep, comparing the alloy of this invention (CM 186 LC) as-cast and double aged, with DS CM 247 LC "super solutioned" and double aged;

Fig. 4A is a group of optical photomicrographs taken away from the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at  $450\text{ MPa}/850^{\circ}\text{C}$  ( $65.27\text{ ksi}/1562^{\circ}\text{F}$ ) after 719.3 hours;

Fig. 4B is a group of optical photomicrographs taken close to the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at  $450\text{ MPa}/850^{\circ}\text{C}$  ( $65.27\text{ ksi}/1562^{\circ}\text{F}$ ) after 719.3 hours;

Fig. 5A is a group of optical photomicrographs taken away from the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at 120 MPa/1050 °C (17.40 ksi/1922 °F) after 327.2 hours;

Fig. 5B is a group of optical photomicrographs taken close to the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at 120 MPa/1050 °C (17.40 ksi/1922 °F) after 327.2 hours;

Fig. 6A is a group of optical photomicrographs taken away from the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at 150 MPa/950 °C (21.76 ksi/1742 °F) after 2985.9 hours;

Fig. 6B is a group of optical photomicrographs taken close to the fracture of the post-test bar (6.35 mm dia.) of the alloy of this invention stressed longitudinally to the grain direction, which failed from creep-rupture at 150 MPa/950 °C (21.76 ksi/1742 °F) after 2985.9 hours;

Fig. 7A is a group of optical photomicrographs taken away from the fracture of the post-test bar (1.78 mm dia.) of the alloy of this invention stressed transversely to the grain direction, which failed from stress-rupture at 172 MPa/982 °C (25.0 ksi/1800 °F) after 442.0 hours;

Fig. 7B is a group of optical photomicrographs taken close to the fracture of the post-test bar (1.78 mm dia.) of the alloy of this invention stressed transversely to the grain direction, which failed from stress-rupture at 172 MPa/982 °C (25.0 ksi/1800 °F) after 442.0 hours;

Fig. 8A is a group of optical photomicrographs taken away from the fracture of the post-test bar (1.78 mm dia.) of the alloy of this invention stressed transversely to the grain direction, which failed from stress-rupture at 138 MPa/1038 °C (20 ksi/1900 °F) after 161.5 hours;

Fig. 8B is a group of optical photomicrographs taken close to the fracture of the post-test bar (1.78 mm dia.) of the alloy of this invention stressed transversely to the grain direction, which failed from stress-rupture at 138 MPa/1038 °C (20 ksi/1900 °F) after 161.5 hours;

Fig. 9 is a photograph of a complex, cooled, multi-airfoil vane segment cast by columnar grain, directional solidification from the alloy of this invention.

In the graphs shown in Figures 1 to 3 the ordinates represent stress in kilograms per square inch and the abscissal represent  $P = (T + 460)[20 + \log 10]10^{-3}$  in which T is the temperature in degrees Fahrenheit and t is the time.

Alloys according to this invention are rhenium containing, directionally solidified (DS) columnar grain, nickel base, superalloys which, in the as-cast and double aged condition, have creep-rupture properties between that of DS CM 247 LC "super solutioned" and double aged (US-A-4461659) and single crystal first generation alloys such as CMSX-2 or 3 (US-A-4582548). In developing such an improved creep strength DS alloy, it was necessary to attain, as much as possible, the high temperature creep resistance of single crystal alloys without resorting to high temperature solution treatment and while also avoiding the grain boundary cracking problems which have been experienced during DS casting of thin wall, cooled airfoils in very high strength alloys.

It has been recognized that zirconium (Zr) has a tendency to migrate to grain boundaries and, consequently, Zr containing superalloys have the characteristic of sensitivity to small changes in Zr content in relation to grain boundary strength, ductility and DS castability. However, the experimental work done in developing this alloy revealed evidence that the phenomena of grain boundary cracking during the DS casting process is extremely sensitive to minute changes in both the Zr and silicon (Si) content of the alloy ingot. Experimentation and testing revealed that reducing these two elements to low levels such as 0.017% for Zr and 0.02% for Si was a partially successful means of overcoming the DS grain boundary cracking problem. In fact, it was found to be very beneficial to further reduce these levels to 0.006% Zr and 0.007% Si. This improvement proved to be quite significant, permitting the most complex, thin wall, cooled vane segments and turbine blades to be produced by DS columnar grain casting without any occurrence of grain boundary cracking during the DS casting operation. Zr has long been used in this type of alloy to increase creep strength. It was found during the experimental work that the reduction in alloying Zr to an extremely low level (0.006%) did not reduce the creep strength but increased the DS transverse rupture ductility. The ultra low Zr nearly doubles the DS transverse 379 MPa/871 °C (55.0 ksi/1600 °F) stress-rupture ductility compared to the earlier CM work.

Fig. 2 graphically shows that VF 714 slab in the as-cast and double aged condition exhibits an appreciable increase in transverse strength over DS CM 247 LC in the "super solutioned" and single aged condition. Further, the DS transverse rupture ductility is improved as evidenced by increasing the elongation from 3.9-4.8% for a heat having .018% Zr to 7.3-10.6% with the super low Zr composition (0.006%) of the VF 714 heat tested at 379 MPa/871 °C (55.0 ksi/1600 °F) (Table IV). The ultra low Zr content (0.006%) of the VF 714 heat appears to have doubled the transverse stress-rupture ductility. The transverse stress-

rupture characteristics of the new alloy are tabulated in Table IV.

Research suggested a reduction in chromium (Cr) and increase in the base element nickel (Ni) was desirable to minimize TCP phase formation. Optimization of tungsten (W) and Re alloy chemistry is helpful in controlling the tendency to form a possible Re-W M<sub>6</sub>C platelet phase or Re-W containing TCP phases following high temperature stress exposure.

As a result of several years of experimenting and testing, an alloy was developed having the ingot composition set out in Table I.

TABLE I

(Wt. % or Wt. ppm)

10	C	.05-.09
	Si	.02 max.
15	Mn	.02 max.
	P	.005 max.
	S	12 ppm max.
20	Cr	5.5-7.0
	<u>Co</u>	9.0-9.5
	Mo	.30-.70
25	W	8.0-9.0
	Ti	.5-.9
	Al	5.5-6.0
	Ta	3.0-4.0
30	Hf	1.2-1.8
	B	.010-.024
	Zr	.004-.010
35	Fe	.15 max.
	Cu	.01 max.
	Nb	.10 max.
40	Mg	80 ppm max.
	V	.05 max.
	Re	2.8-3.1
45	[N]ppm	10 max.
	[O]ppm	10 max.
	Ni	Balance.

50 Based upon this alloy specification, heat VF 714 was prepared having the specific composition set out in Table II.

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TABLE II  
(Wt. % or Wt. ppm)

5	C	.070
	Si	.007
	Mn	.003
	Cr	6.6
10	Mo	.50
	Fe	.11
	P	<.005
15	S ppm	3
	Cu	<.001
	Co	9.2
	V	<.005
20	Nb	<.05
	Ta	3.2
	W	8.5
25	Al	5.68
	Ti	.69
	B	.016
30	Re	3.0
	Hf	1.40
	Zr	.006
35	Mg ppm	13
	[N] ppm	2
	[O] ppm	3
	Ni	Balance

40 Specimens were machined from a slab DS cast from the VF 714 heat. The specimens were, as-cast and double aged heat treated, as described in Tables III and IV and then tested longitudinal and transverse to the grain direction for creep- and stress-rupture properties. The results of these tests are set out in Tables III and IV.

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TABLE III

CM 186 LC (VF 714) (Ultra Low Zr)  
 DS Slab  
 DS Longitudinal Creep-Rupture  
 As-Cast + Double Aged [4 hrs/1080°C (1975°F) AC +  
 20 Hrs/1600°F (871°C) AC] 6.35 mm Dia (0.250")  
 Gage Solid Specimens.

Specimen #	Test Conditions	% Plastic Strain at 20 hrs	Time (Hrs) To Creep		Rupture Life Hrs	Elong. % 4D	RA %
			1.0%	2.0%			
VF714 1L	450 MPa/850°C (65.27 ksi/1562°F)	0.169	169.1	284.3	719.3	23.7	30.0
VF714 2L	150 MPa/950°C (21.76 ksi/1742°F)	0.018	1300.1	1810.2	2985.9	40.2	49.2
VF714 3L	120 MPa/1050°C (17.40 ksi/1922°F)	0.383	120.7	212.8	327.2	36.0	55.4

TABLE IV

CM 186 LC (VF 714) (Ultra Low Zr)  
 DS Slab  
 DS Transverse Stress-Rupture  
 As-Cast + Double Aged [4 Hrs/1080°C (1975°F) AC +  
 20 Hrs/871°C (1600°F) AC] 1.78 mm Dia. (0.070")  
 Gage Solid Specimens

Specimen #	Test Conditions	Rupture Life Hrs	Elong. % 4D	RA %
VF714 B1T	379 MPa/871°C (55.0 ksi/1600°F)	413.2	10.6	19.6
VF714 T4T	379 MPa/871°C (55.0 ksi/1600°F)	417.6	7.3	11.7
VF714 B2T	172 MPa/982°C (25.0 ksi/1800°F)	442.0	8.3	20.4
VF714 T5T	172 MPa/982°C (25.0 ksi/1800°F)	258.9	8.8	15.7
VF714 B3T	138 MPa/1038°C (20.0 ksi/1900°F)	161.5	13.9	20.6
VF714 T6T	138 MPa/1038°C (20.0 ksi/1900°F)	96.8	11.2	18.9

The results of these tests are graphically summarized in the Larson-Miller curves in Figs. 1, 2 and 3. The results in Fig. 1 graphically compare the alloys DS CM 247 LC and CMSX-2/3 and the alloy of this invention DS CM 186 LC. The graphical data of Figs. 2 and 3 compare DS CM 247 LC and the alloy of this invention. The data point in Fig. 2 at approximately 55 ksi/P = 46.6 is a dual plot because of the identical values for two of the points. No comparison is made in Figs. 2 and 3 with CMSX-2 or 3 alloys because no comparative data is available. All three graphs show a marked improvement in the functional characteristics of creep- and stress-rupture properties of DS CM 186 LC as-cast and double aged over DS CM 247 LC, which was in the "super solutioned" and double or single aged condition. The results graphically set out in Fig. 1 make it clear that this alloy is a marked improvement over DS CM 247 LC without having to resort to the creep strengthening high temperature solution treatment and provides an alloy having functional qualities that are closer to the single crystal alloys.

A very severe creep-rupture test was conducted on a slab from the VF 714 heat in the as-cast and double aged condition. This test continued at 150 MPa/950°C (21.76 ksi/1742°F) for 2985.9 hours before rupture. This result was not plotted on the Larson-Miller curve (Fig. 1) because no even remotely comparable data for DS CM 247 LC or for CMSX-2/3 in terms of hours of life under this type of stress/temperature conditions is available.

A limited number of both IP (intermediate pressure) turbine blades and HP (high pressure) turbine blades have been DS cast from heat VF 714, using production process procedures. The IP blades were cast to determine any core to metal reaction problems and none was found. The HP cored turbine blades were cast root-down and convex-out, since the convex-out attitude of this component has a history of being prone to DS grain boundary cracking and is, therefore, a stiff test for an alloy's DS castability. No DS grain boundary cracking problems were encountered with heat VF 714.

Two IP and two HP blades were extensively sectioned to a turbine engine company specification cut-up plan and examined for levels of microporosity using a Quantimet 720 image analyzer. The blade porosity generally satisfied the <0.5% background figure and the isolated field criteria. Also, no continuous grain boundary carbide or gamma/gamma prime eutectic networks were apparent.

DS grain boundary cracking problems were encountered during casting an advanced, complex vane segment with an earlier version of the CM 186 LC alloy using CM heat VF 686 (0.02% Zr and 0.02% Si). CM heat VF 714 with ultra low Zr and Si contents (0.006% Zr and 0.007% Si) was successful in overcoming the DS grain boundary cracking problem inherent with the earlier heat VF 686, when DS cast into the advanced, complex vane segment. This component represents an extremely severe DS castability test for any high strength cast superalloy.

It will be recognized that this invention provides improved creep- and stress-rupture strength DS alloys that are usable under at least two different casting and application circumstances for which no heretofore known alloy has been eminently suitable. It provides alloys capable of sustained use as LP turbine blades designed for high temperature/high stress environments. These blades for large turbofan engines have to be of a length such that present single crystal technology is not capable of or practical for producing. It also provides alloys for use in turbine vane segments designed for use in high temperature/high stress environments in which the necessary shape of the vane segment, such as that illustrated in Fig. 9, makes use of a single crystal casting and full gamma prime solution treatment impossible, without occurrence of deleterious casting grain defects and recrystallization.

The composition of the alloys of the invention indicate they would be tolerant to the addition of very small amounts of one of the rare earth elements including yttrium (Y), lanthanum (La) or cerium (Ce). This addition would be in a very restricted range of 0.2% or less of the combined total of these elements in the ingot. This total can consist of only one of these elements or it can consist of two or three of these elements. Since these are very reactive elements, the alloy as DS cast into a finished component would probably contain only about 100 to 150 ppm of the rare earth element. However, these elements are highly effective in improving bare oxidation performance and even this small amount has a marked beneficial effect on bare oxidation resistance. Further, since this alloy contains appreciable hafnium (Hf) which is a known "protector" of rare earth elements and the DS casting temperature of this alloy is substantially less than that of single crystal alloys, there will be less reaction between the rare earth elements and the ceramics of the mold or core. The Hf has a propensity for forming a protective hafnium oxide film between the solidifying alloy and the mold or core ceramic materials.

The invention provides directionally solidified alloys capable of being DS columnar grain cast under the foundry sophistication of production facilities experienced in casting aircraft turbine engine blades and vane segments with dependable and repeatable quality results. It provides alloys capable of use in the casting of turbine blades and vanes, the operating conditions of which are beyond the temperature and stress capabilities of previously known, directionally solidified alloys such as CM 247 LC, yet the casting configuration makes the use of single crystal alloy technology either impractical or impossible. This is achieved without the resort to high temperature solution treatment for creep strengthening. Therefore, it provides a solution to a presently increasingly difficult and demanding situation. The invention addresses the further problem that, as the necessity of increased performance and efficiency of aircraft engines becomes both desirable and necessary, an increasingly serious technical barrier has been encountered between the need for complex turbine vane segments and for low pressure turbine blades of length such that there is a lack of availability of superalloys from which they can be cast. The invention provides a solution to this problem by providing alloys capable of the high temperature performance required of these engines yet capable of being DS cast and used in the as-cast and double aged condition.

Recently acquired data from the continuing testing and evaluation of an alloy of this invention with ultra low Zr indicates excellent DS transverse stress-rupture properties such that the alloy is expected to demonstrate improved equiaxed stress- and creep-rupture properties over the CM 247 LC alloy. Thus, it is expected to have use in conventionally cast products such as integral turbine wheel or turbocharger rotors in small gas turbines and high performance piston or rotary engines where the use involves high stress and temperature conditions.

Having described the chemistry of the new alloy and its novel functional characteristics and the problems it addresses and solves in the new generation of larger and more complex, high performance aircraft turbofan LP turbine blades and HP and IP vane segments, it is submitted that Applicant is entitled to protection for both the alloy and turbine blades and vane segments cast therefrom. It is also envisaged that the alloys may find DS turbine blade applications in large industrial gas turbines using clean fuels such as natural gas. The blade lengths in these large industrial turbines can range from 305 to 610 mm (12" to 24") long.

While the invention has been described primarily in relation to alloys having high temperature creep resistance for use in directionally solidified, columnar grain turbine blades and turbine vanes, alloys in accordance with the invention may have other uses. One such use is for a conventionally cast, equiaxed grain, polycrystal, integral turbine wheel or turbocharger rotor for use under high stress, high temperature conditions in small gas turbines and high performance piston and rotary engines.

### Claims

1. A low carbon nickel base alloy consisting of the following elements (Wt. % or ppm):

C	.05-.09
Si	.02 max.
Mn	.02 max.
P	.005 max.
S	12 ppm max.
Cr	5.5-7.0 <i>low</i>
Co	9.0-9.5 <i>high</i>
Mo	.30-.70 <i>low</i>
W	8.0-9.0 <i>high</i>
Ti	.5-.9
Al	5.5-6.0 <i>high</i>
Ta	3.0-4.0 <i>low</i>
Hf	1.2-1.8 <i>high</i>
B	.010-.024
Zr	.004-.010
Fe	.15 max.
Cu	.01 max.
Nb	.10 max.
Mg	80 ppm max.
V	.05 max.
Re	2.8-3.1 ✓
[N]ppm	10 max.
[O]ppm	10 max.
and optionally, from 0 to 0.2% of a combination of one, two or three of Y, La and Ce	
Ni	Balance.

2. An alloy according to claim 1 consisting of the following elements (Wt. % or ppm):

5	C	.070
	Si	.007
	Mn	.003
	Cr	6.6
	Ni	Balance
	Mo	.50
	Fe	.11
	P	<.005
10	S ppm	3
	Cu	<.001
	Co	9.2
	V	<.005
	Nb	<.05
15	Ta	3.2
	W	8.5
	Al	5.68
	Ti	.69
	B	.016
20	Re	3.0
	Hf	1.40
	Zr	.006
	Mg ppm	13
	[N] ppm	2
25	[O] ppm	3

3. An alloy according to claim 1 or claim 2 which is essentially free of excessive M<sub>6</sub>C or TCP phases after continuous exposure to 150MPa at 950 °C for a minimum of 2900 hours.
4. A method of making a directionally solidified airfoil with a columnar grain structure which comprises casting the airfoil from an alloy according to claim 1 or claim 2 or claim 3 followed by ageing for four hours at 1080 °C, then directly gas cooling, then holding for twenty hours at 871 °C then directly gas cooling.
5. A vacuum cast directionally solidified, columnar grain component made from an alloy according to claim 1 or claim 2 or claim 3 characterised by a configuration which cannot be practically cast as a substantially grain defect free single crystal.
6. The use of an alloy according to claim 1 or claim 2 or claim 3 for the manufacture as a directionally solidified, columnar grain casting of a turbine blade or vane for use under high stress, high temperature conditions.
7. The use of an alloy according to claim 1 or claim 2 or claim 3 to produce a conventionally cast, equiaxed grain, polycrystal, integral turbine wheel or turbocharger rotor for use under high stress, high temperature conditions in small gas turbines and high performance piston and rotary engines.

#### Patentansprüche

1. Legierung auf Nickelbasis mit niedrigem Kohlenstoffgehalt bestehend aus den folgenden Elementen (Gew.-% oder ppm):

5	C	.05-.09
	Si	.02 max.
	Mn	.02 max.
	P	.005 max.
	S	12 ppm max.
	Cr	5.5-7.0
	Co	9.0-9.5
	Mo	.30-.70
10	W	8.0-9.0
	Ti	.5-.9
	Al	5.5-6.0
	Ta	3.0-4.0
	Hf	1.2-1.8
15	B	.010-.024
	Zr	.004-.010
	Fe	.15 max.
	Cu	.01 max.
	Nb	.10 max.
20	Mg	80 ppm max.
	V	.05 max.
	Re	2.8-3.1
	[N]ppm	10 max.
	[O]ppm	10 max.
25	und fakultativ von 0 bis 0,2 % einer Kombination von einem, zwei oder drei von Y, La und Ce mit	
	Ni	Ausgleich.

30 2. Legierung nach Anspruch 1, bestehend aus den folgenden Elementen (Gew.% oder ppm):

35	C	.070
	Si	.007
	Mn	.003
	Cr	6.6
	Ni	Ausgleich
	Mo	.50
	Fe	.11
40	P	<.005
	S ppm	3
	Cu	<.001
	Co	9.2
	V	<.005
45	Nb	<.05
	Ta	3.2
	W	8.5
	Al	5.68
	Ti	.69
50	B	.016
	Re	3.0
	Hf	1.40
	Zr	.006
	Mg ppm	13
55	[N] ppm	2
	[O] ppm	3

3. Legierung nach Anspruch 1 oder 2, welche im wesentlichen frei ist von übermäßigen M<sub>6</sub>C- oder TCP-Phasen nach der stetigen Beaufschlagung unter 150MPa bei 950 °C für mindestens 2900 Stunden.
4. Verfahren zum Herstellen eines gerichtet erstarrten Flügels mit einer säulenartigen Kornstruktur, welches das Gießen des Flügels aus einer Legierung gemäß einem der Ansprüche 1 oder 2 oder 3 mit anschließendem Altern für vier Stunden bei 1080 °C, dann direktem Gaskühlen, dann Halten bei 871 °C für 20 Stunden, dann direktem Gaskühlen umfaßt.
5. Vakuumgeformte gerichtet erstarrte Komponente mit säulenartigem Korn, die aus einer Legierung gemäß Anspruch 1, 2 oder 3 hergestellt wurde, gekennzeichnet durch eine Konfiguration, welche praktisch nicht als ein im wesentlichen korndefektfreier Einkristall gegossen werden kann.
6. Verwendung einer Legierung nach den Ansprüchen 1, 2 oder 3 zur Herstellung eines gerichtet erstarrten Gußteils mit säulenartigem Korn einer Turbinenschaufel oder eines Turbinenflügels zur Verwendung unter hohen Belastungs- und Temperaturbedingungen.
7. Verwendung einer Legierung nach den Ansprüchen 1, 2 oder 3 zur Herstellung eines konventionell gegossenen ein gleichartig gerichtetes Korn aufweisenden, polykristallinen, integralen Turbinenrades oder Turbolader-Rotors zur Verwendung unter hohen Belastungs- und Temperaturbedingungen in kleinen Gasturbinen und Hochleistungs-Kolben- oder Rotormaschinen.

#### Revendications

1. Alliage à base de nickel contenant peu de carbone, consistant en les éléments suivants (% en poids ou ppm) :

C	0,05 à 0,09
Si	max 0,02
Mn	max 0,02
P	max 0,005
S	max 12 ppm
Cr	5,5 à 7,0
Co	9,0 à 9,5
Mo	0,30 à 0,70
W	8,0 à 9,0
Ti	0,5 à 0,9
Al	5,5 à 6,0
Ta	3,0 à 4,0
Hf	1,2 à 1,8
B	0,010 à 0,024
Zr	0,004 à 0,010
Fe	max 0,15
Cu	max 0,01
Nb	max 0,10
Mg	max 80 ppm
V	max 0,05
Re	2,8 à 3,1
[N]ppm	max 10
[O]ppm	max 10
et, facultativement, de 0 à 0,2% d'une combinaison de un, deux ou trois des Y, La et Ce,	
Ni	complément.

2. Alliage suivant la revendication 1, consistant en les éléments suivants (% en poids ou ppm) :

5

10

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C	0,070
Si	0,007
Mn	0,003
Cr	6,6
Ni	complément
Mo	0,50
Fe	0,11
P	< 0,005
S ppm	3
Cu	< 0,001
Co	9,2
V	< 0,005
Nb	< 0,05
Ta	3,2
W	8,5
Al	5,68
Ti	0,69
B	0,016
Re	3,0
Hf	1,40
Zr	0,006
Mg ppm	13
[N]ppm	2
[O]ppm	3

30

35

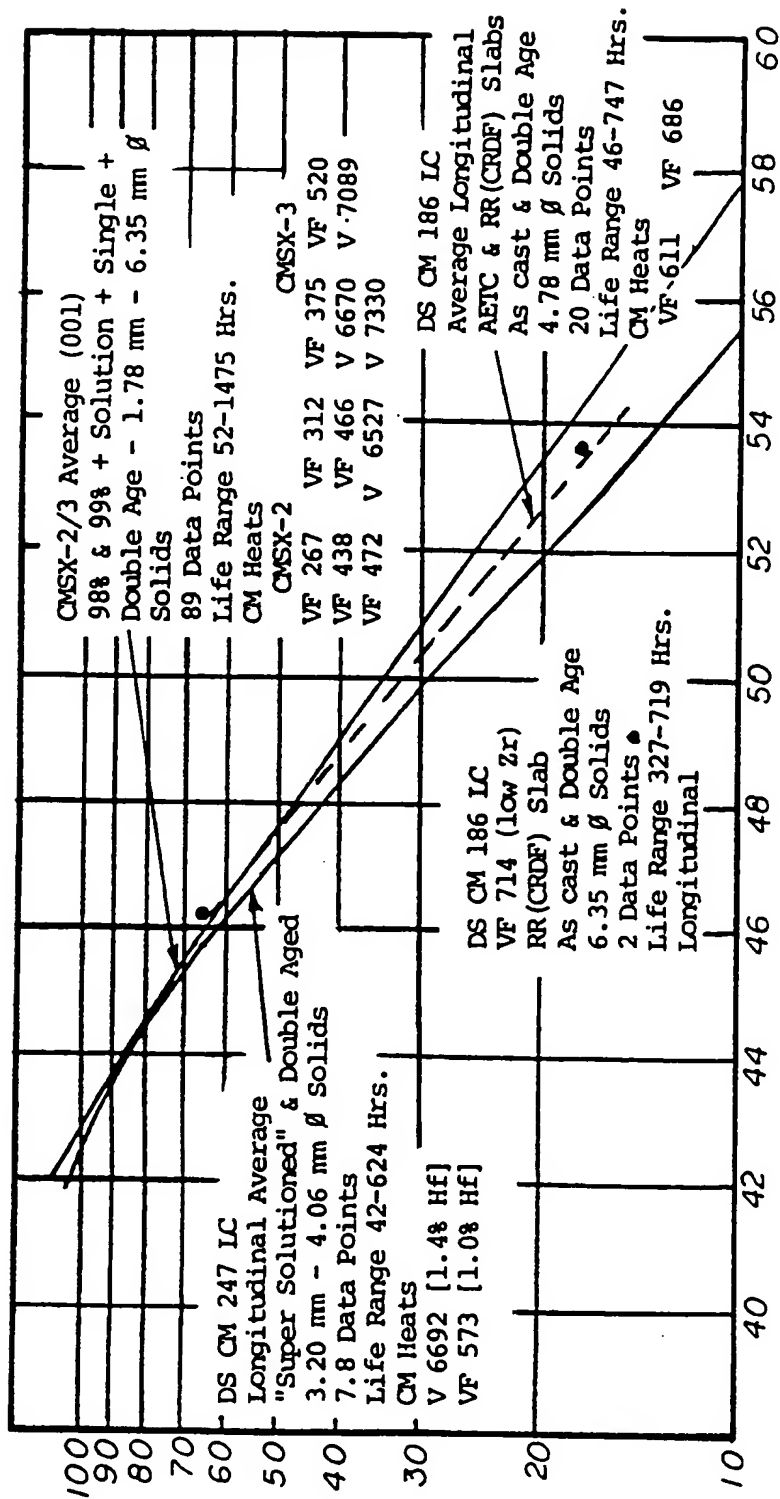
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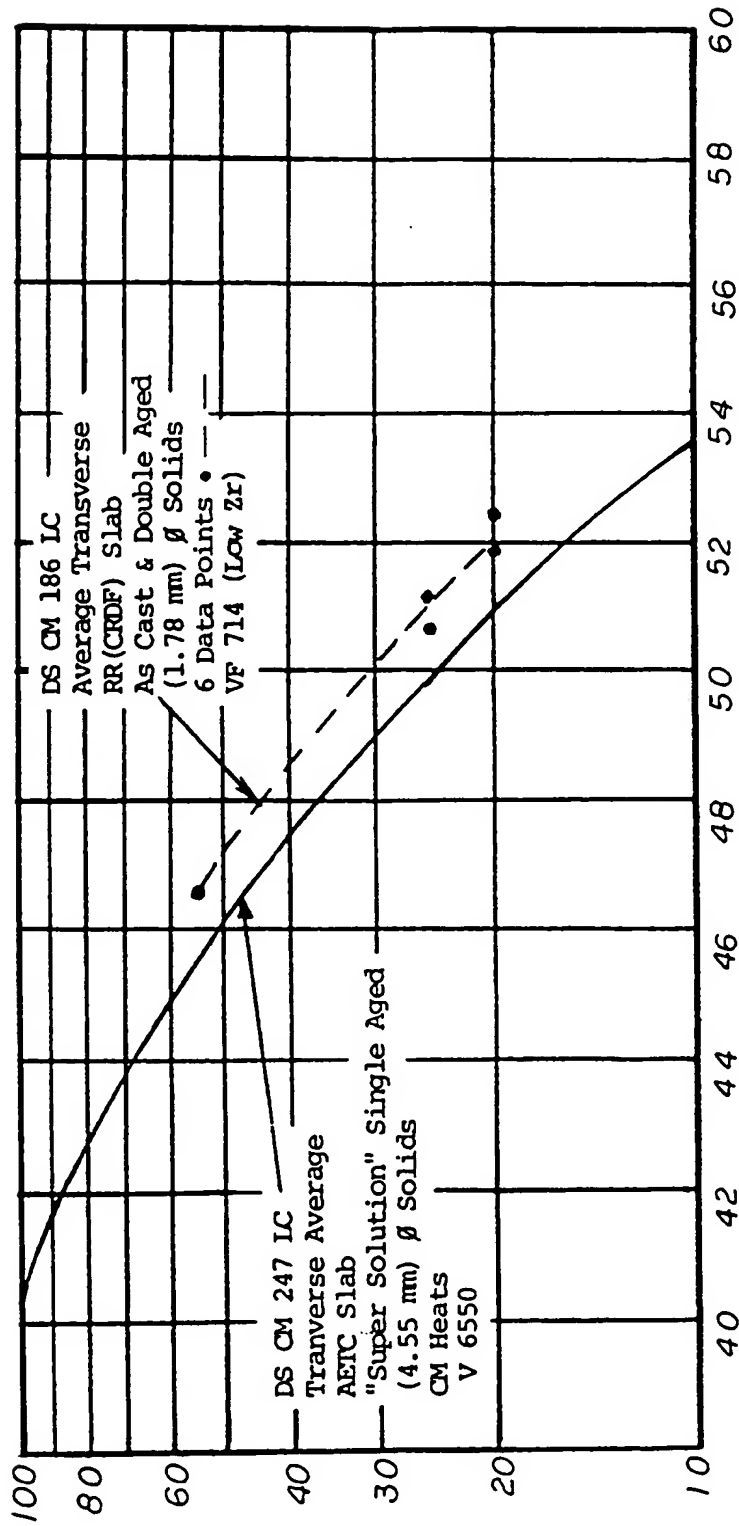
3. Alliage suivant la revendication 1 ou 2, qui est essentiellement exempt d'un excès de phases M6C ou TCP, après exposition continue à 150 MPa à 950 °C pendant un minimum de 2900 heures.
4. Procédé de réalisation d'une surface portante à solidification orientée avec une structure en grains colonnaires, qui comprend le moulage de la structure portante à partir d'un alliage suivant la revendication 1, 2 ou 3, puis par vieillissement pendant 4 heures à 1080 °C, puis, directement, refroidissement par gaz, ensuite maintien pendant 20 heures à 871 °C, ensuite, directement refroidissement par gaz.
5. Composant en grains colonnaires, à solidification orientée, moulé sous vide, constitué d'un alliage suivant la revendication 1, 2 ou 3, caractérisé par une configuration qui ne peut pratiquement pas être moulée comme un monocristal sensiblement exempt de défauts de grains.
6. Utilisation d'un alliage suivant la revendication 1, 2 ou 3, dans la préparation d'un moulage en grains colonnaires, à solidification orientée, d'une lame ou pale de turbine, à utiliser dans des conditions de tension élevée et de température élevée.
7. Utilisation d'un alliage suivant la revendication 1, 2 ou 3, pour produire une roue de turbine ou un rotor de turbochargeur solidaire, polycristallin, en grains équiaxiaux, classiquement moulé, à utiliser dans des conditions de tension élevée et de température élevée, dans des petites turbines à gaz et de moteurs rotatifs et à piston de haute performance.



$$P = (T+460) [20 + \log 10^t] 10^{-3}$$

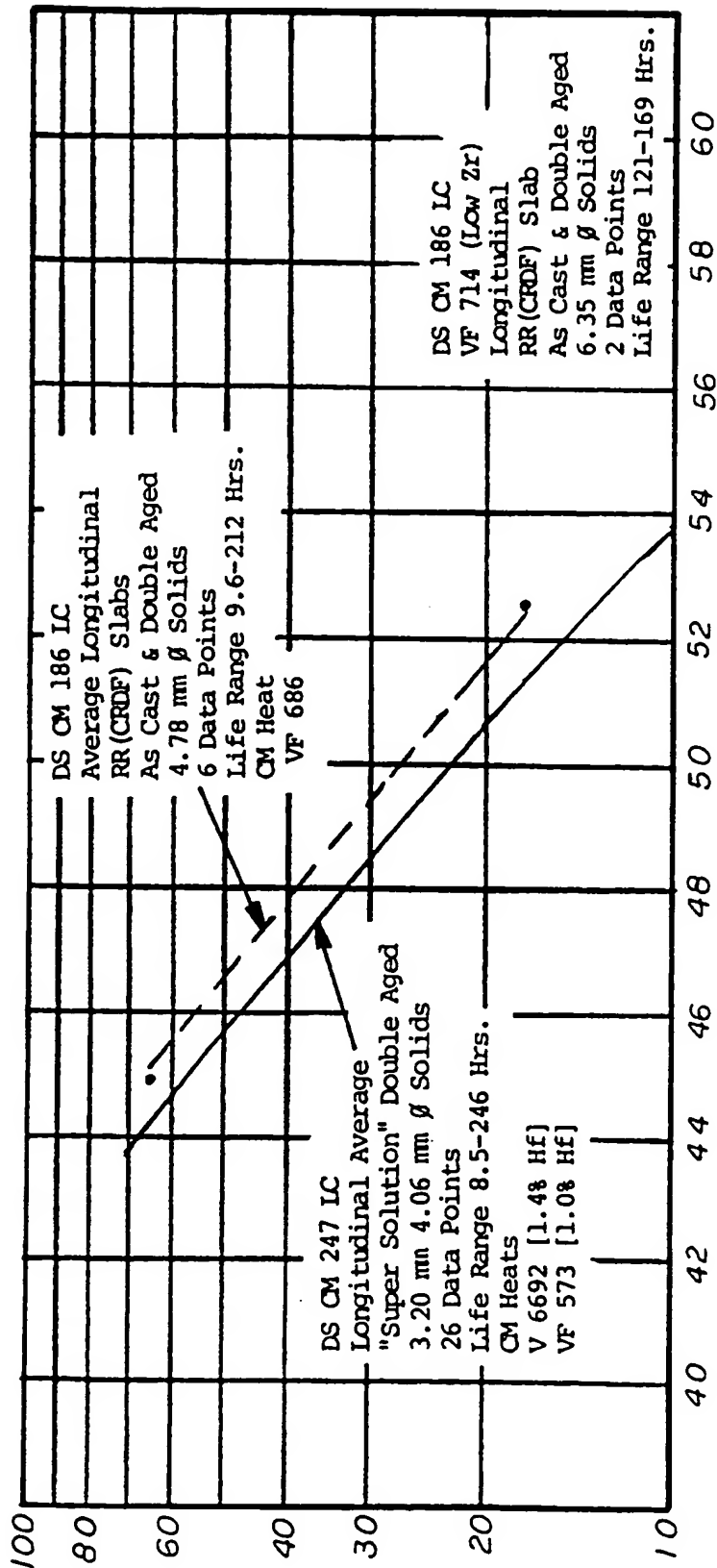
Fig.1





$$p^{44} \propto (T+460) [20 + \log 10^t] 10^{-3}$$

Fig. 2



$$P = (T+460) [20 + \log 10^t] 10^{-3}$$

Fig. 3

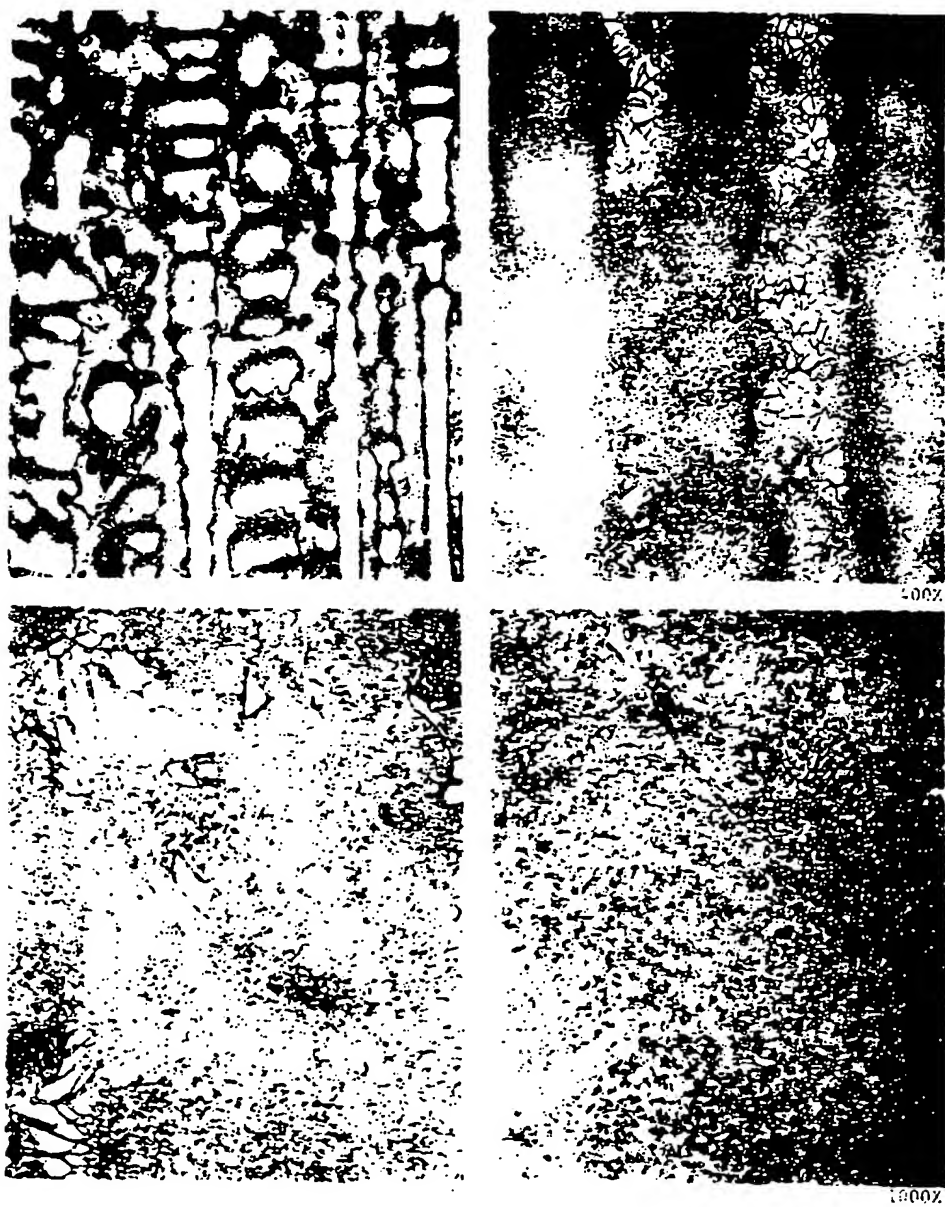


FIG. 4A

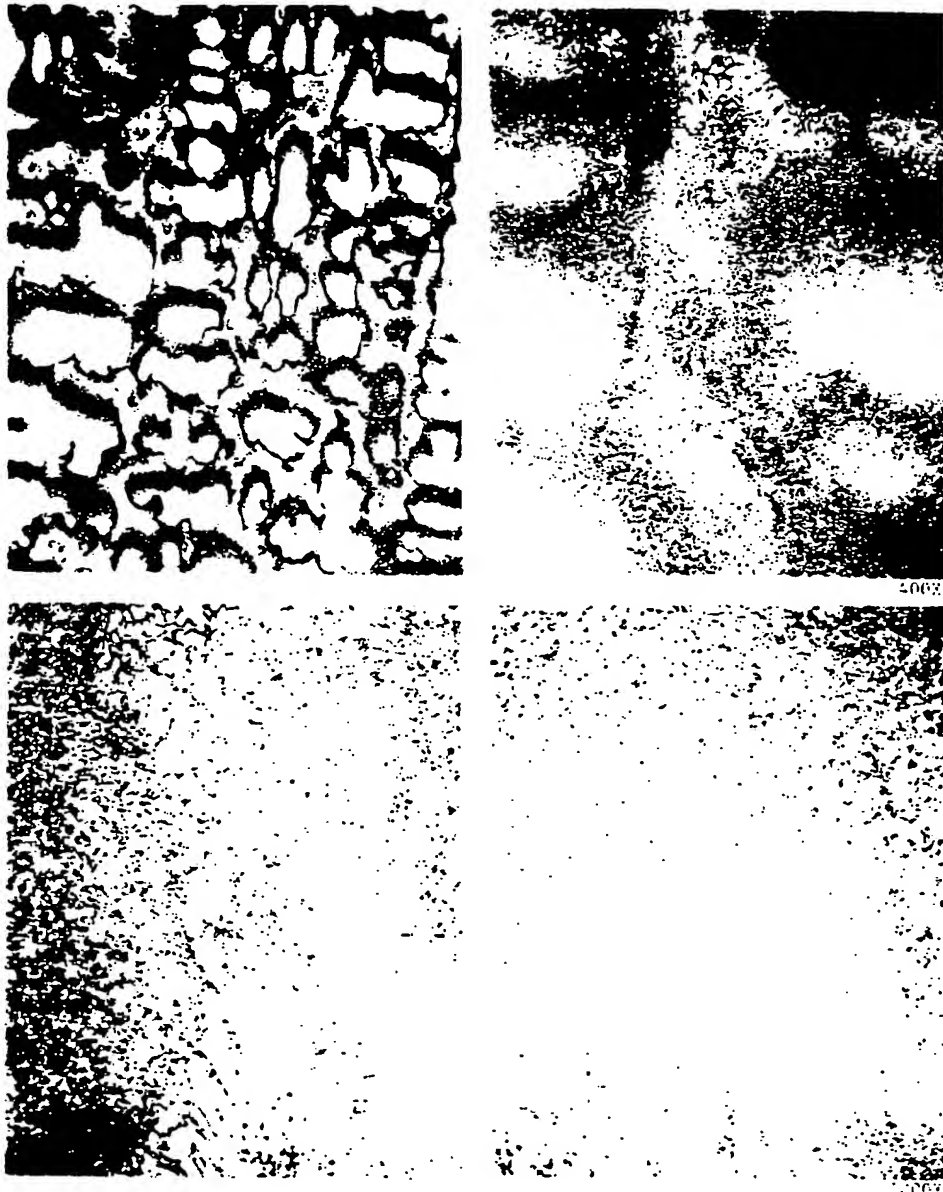


FIG. 4B

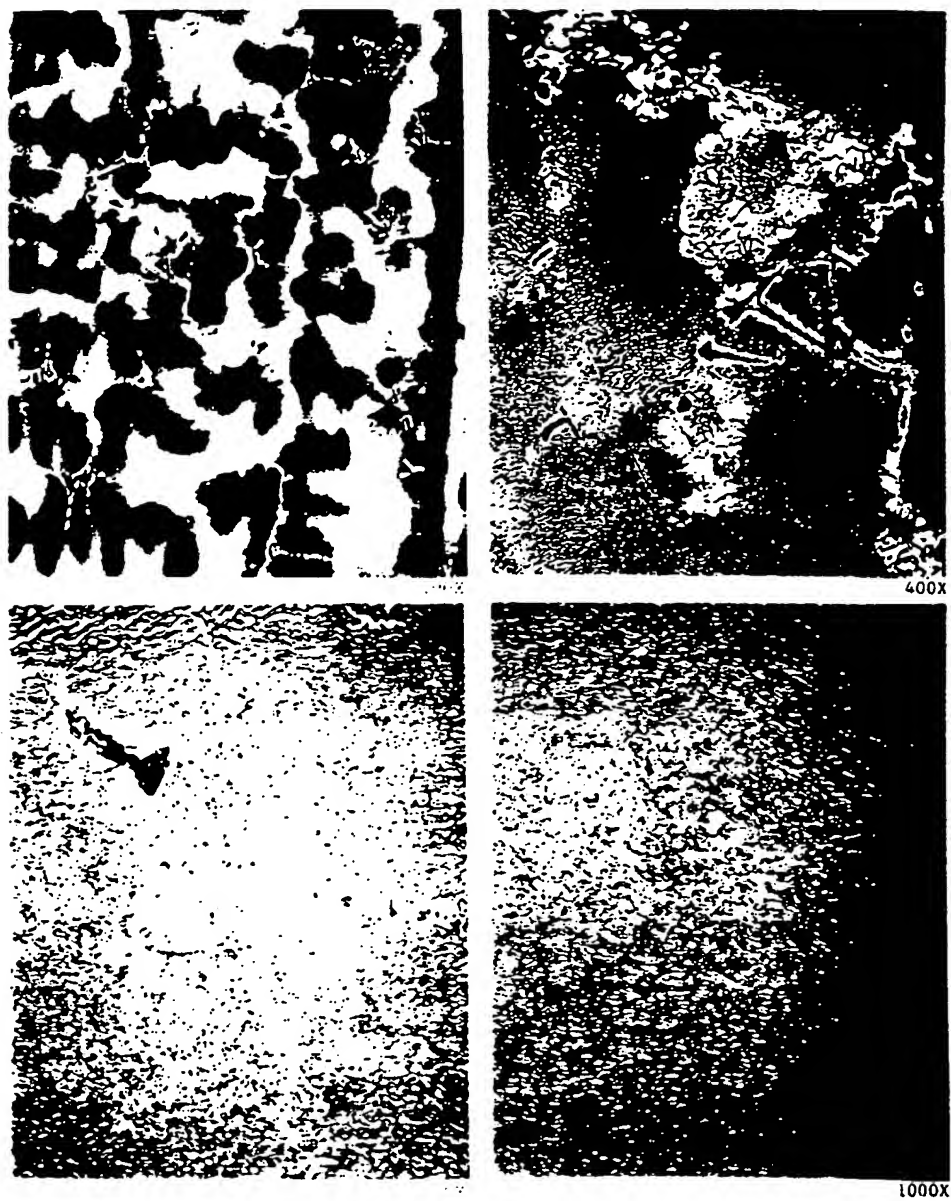


FIG. 5A

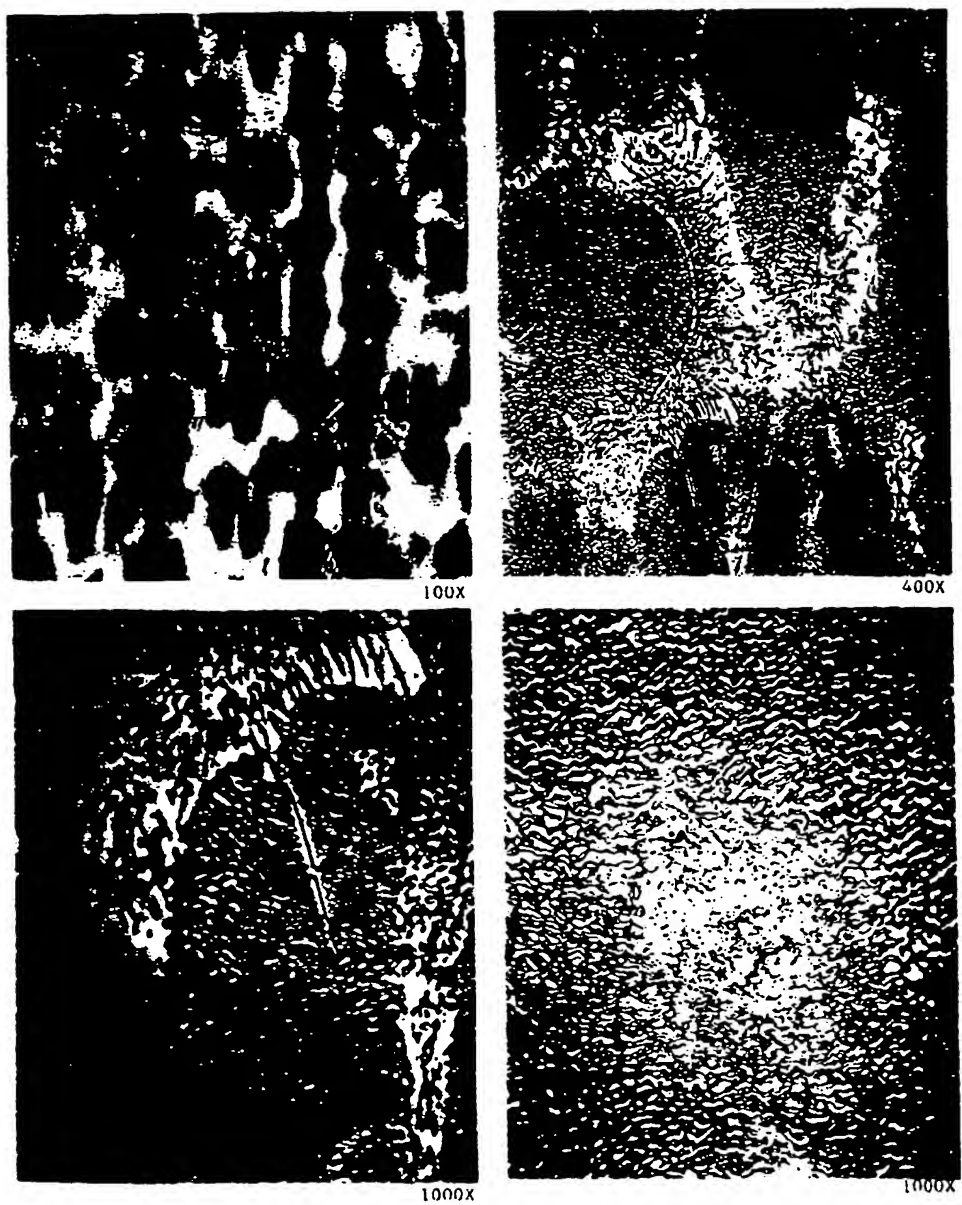


FIG. 5B

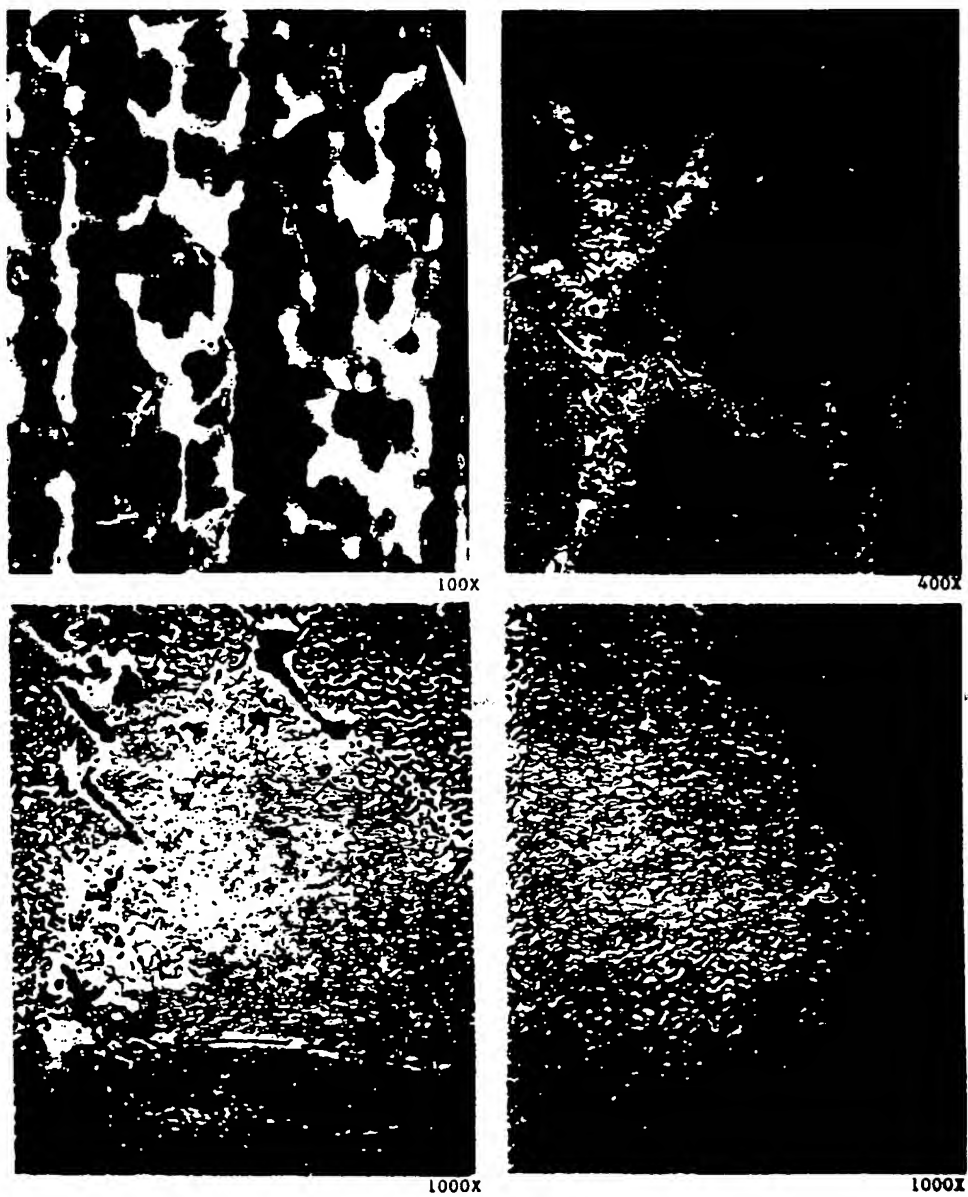


FIG. 6A

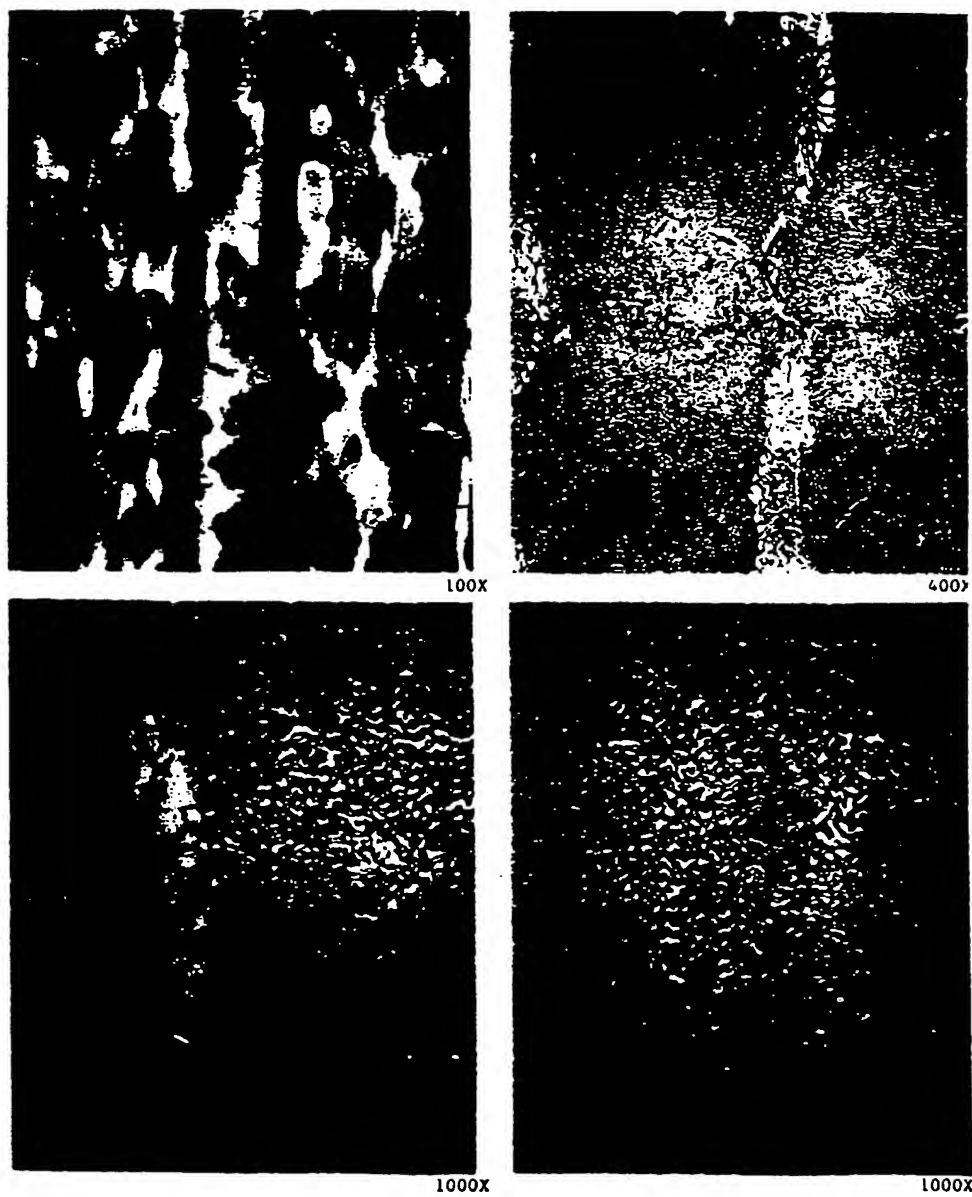


FIG. 6B



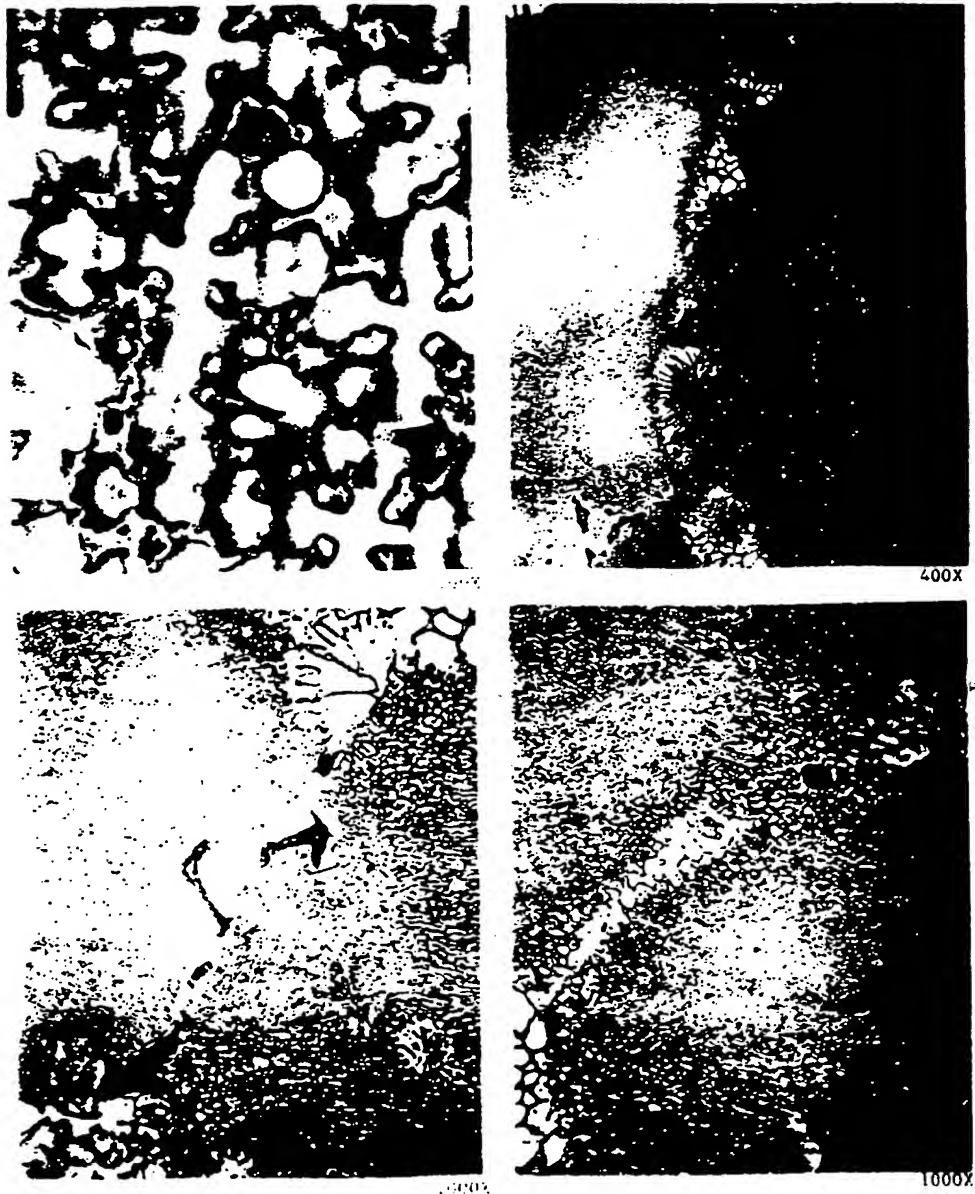


FIG. 7A

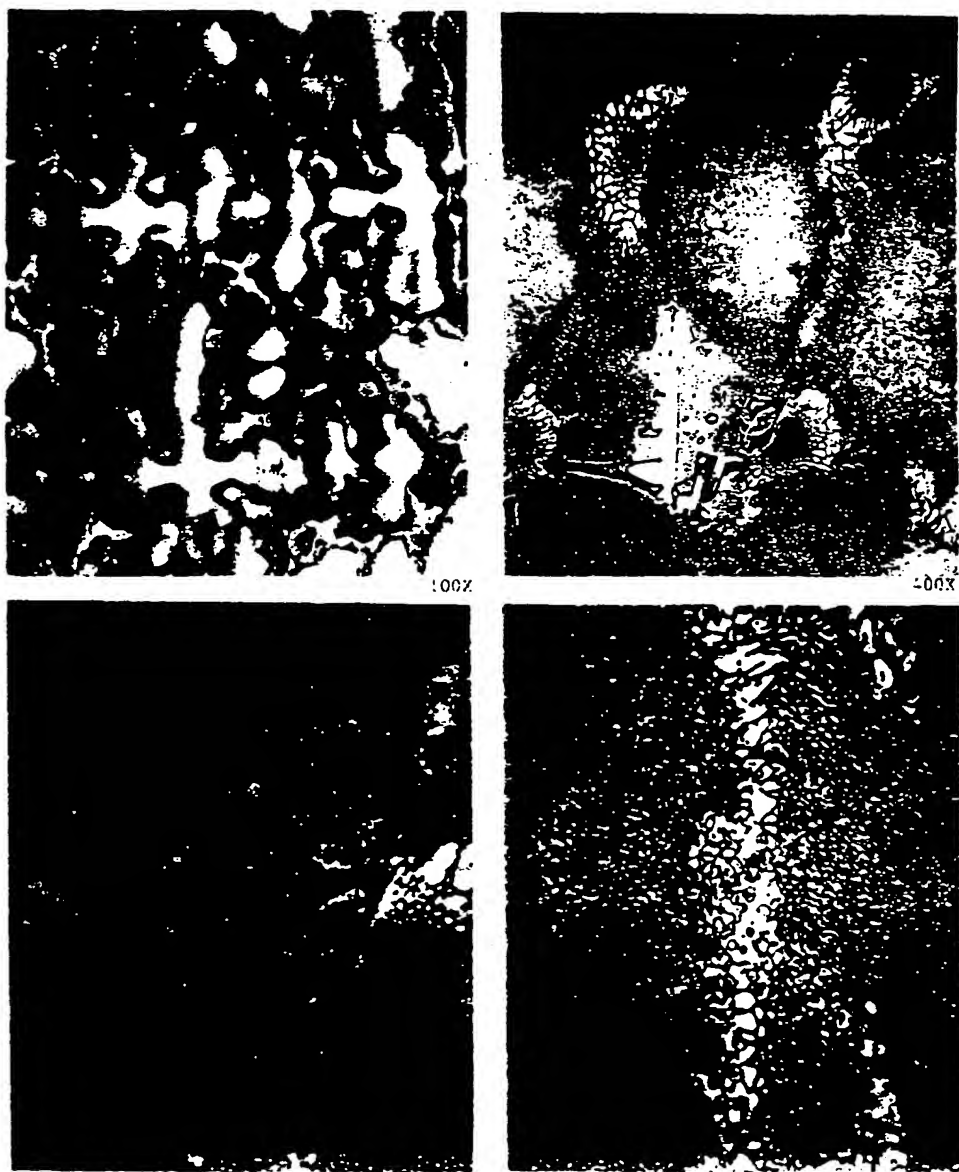


FIG. 7B

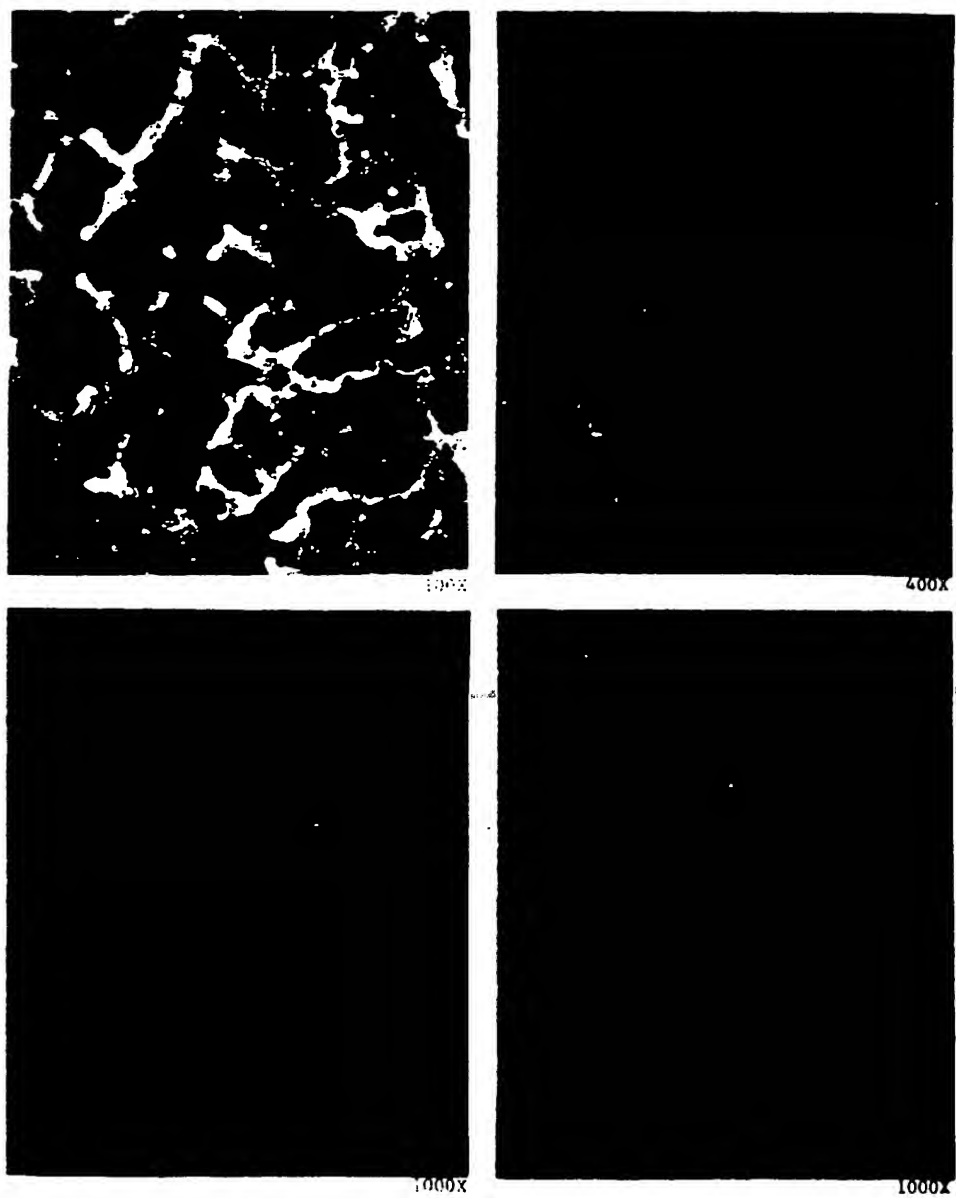


FIG. 8A

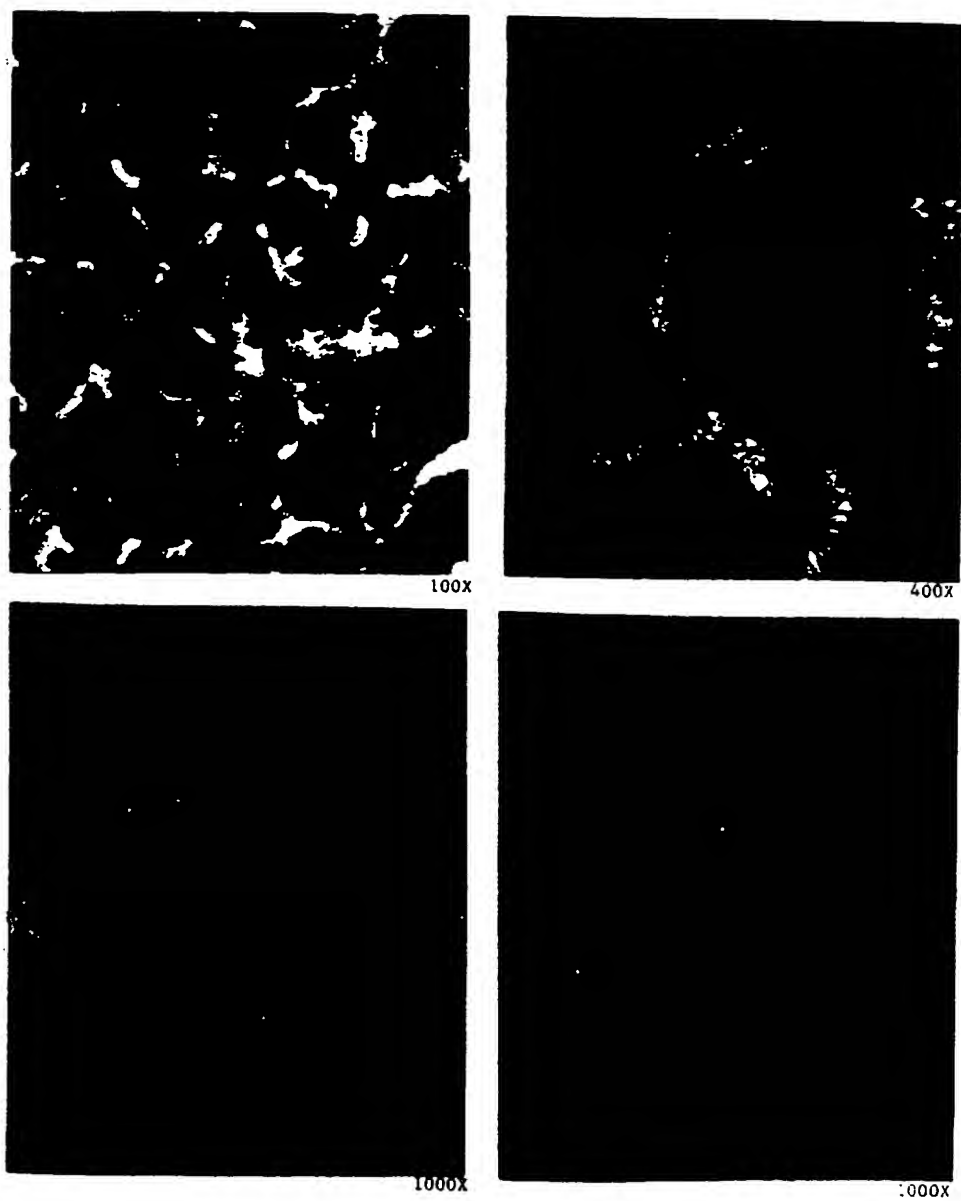


FIG. 8B

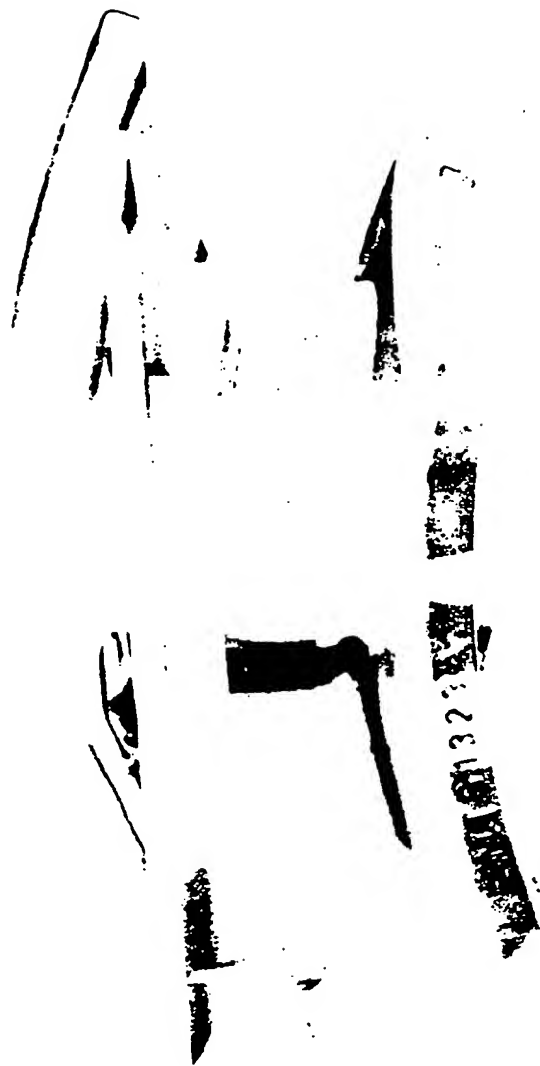


FIG. 9

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